

Microscopical Mechanisms of Crack Growth

V. I. Vladimirov, V. I. Smirnov

A. F. Ioffe Physico - Technical Institute Leningrad
Academy of Sciences of the USSR

1.

It has been shown^{/1/} that strength and energy criteria of crack nucleation caused by the linking of dislocations are not identical. In a number of dislocation models crack nucleation occurs due to thermal agitation at local stresses less than the theoretical strength limit nearly by the order of magnitude. Similar situation is also peculiar to the crack growth. The energy criterion $\frac{\partial W}{\partial h} = w' \ll 0$, W being the system free energy, h - crack length, only means that crack growth is thermodynamically favourable and holds at $\sigma > \sigma_w$ less than the strength criterion $\sigma > \sigma_{th}$, which requires the theoretical strength limit to be reached at the crack tip, $\sigma_w < \sigma_{th}$. The equality $\sigma_w = \sigma_{th}$ is fulfilled for brittle cracks only that are unstable after reaching the Griffith length. It can be easily shown by increasing the curvature radius r of the constant length crack $h > h_c = \text{Const}$, where h_c is the Griffith length: $w'(h_c) = 0$. Then the stresses at the crack tip decrease, $\sigma_i \sim r^{-1/2}$, whereas the Griffith condition does not change.

2.

If $h > h_c$, but $r \gg a$, a , being the interatomic distance (this condition is usually satisfied for all cracks in ductile fracture), the crack growth is not forbidden by the thermodynamical Griffith condition, but as $\sigma_i \ll \sigma_{th}$ at the crack tip one should to consider here the micromechanisms controlling the crack growth. On the other hand it is seen that the Griffith length $h = h_c$ for the crack with $r \gg a$ is no longer the stability threshold. It is clear from $\sigma_i \ll \sigma_{th}$ that even for the case $h > h_c$ the crack will slowly propagate becoming unstable only at $h > h_{\sigma} \approx h_c \cdot \left(\frac{r}{a}\right)^{1/2}$.

However, evidence is available which shows that the Griffith length is an important parameter of the theory. While passing from $h < h_c$ to $h > h_c$ the microscopic mechanisms of crack growth change. Indeed, the growth of an isolated crack with $h < h_c$ is not energetically favourable and requires an influx of energy into the crack from outside. Hence, at $h < h_c$ plastic flow aids in crack growth. On the contrary, $h > h_c$ requires a constant energy flux from the crack so that to make it quasi-stable. If this flux is impossible or insufficient, the crack becomes unstable and brittle fracture occurs.

3.

The table of possible microscopic mechanisms given below is based on the results obtained. Macrocrack velocity for the majority of these mechanisms (except the brittle and plastic ones) is mainly determined by microcrack nucleation in front of the macrocrack. Calculations similar to [1] show that the microcrack nucleation occurs most easily when several dislocations with non-zero Burgers vector unite. This linking becomes possible at the local stresses as low as $\sigma_i \approx 0.1 \sigma_{th}$ and the activation energy for the process does not exceed $G b^3$ by the order of magnitude, G is the shear modulus, b Burgers vector, which is of some ev for metals.

4.

Complex-stress fracture occurs under the action of both external and internal orientated stresses. Assume that the deformation begins near the surface and the slip band penetrates the material by the depth L_n comparable with the scale of inhomogeneity by the order of magnitude, Fig.1. Then dislocations of the same sign came out of the crystal forming a jog. The structure obtained can be described as a superposition of effective dislocations and the dipole structure, Fig.2. The number of effective dislocations is equal to the number of dislocations that came out of the crystal, their signs being opposite. In the calculations the effective dislocations can be consi-

dered as straight lines and their length can be taken as comparable to the scale of inhomogeneity by the order of magnitude. Again the group of effective dislocations can be substituted by one dislocation with $b_e = nb$. Dislocations thus introduced determine the level and the character of internal stresses, as well as the conditions and orientation of the dipole microcrack nucleated. According to Fig.1 and the expression for the plastic strain $\epsilon = \rho_m \cdot b \ell$ (ρ_m is the mobile dislocation density, ℓ - dislocation mean free path length) we get the effective dislocation mean density: $\rho_{ey} = \epsilon \cdot (b L_n)^{-1}$. Generalization for the case when ρ_m and ℓ near the surface differ from those in the volume is quite clear. Put for example, $L_n = 1 \text{ cm}$, $\epsilon = 0.1$. The the effective dislocations at the volume boundary create the stresses $\sigma_i \approx \frac{G}{2\pi} \cdot \frac{\rho_{ey} \cdot L_n^2}{L_n} \cdot b = 100 \text{ kg per mm}^2$.

5.

The concept of effective dislocations was applied to the study of plastic deformation and fracture in rolling. Theoretically and experimentally were determined the systems of effective dislocations, the stress tensor due to dislocations, the conditions of the nucleation and the growth of cracks and the crack orientation. Of utmost interest are the results that are concerned with cross-rolling. Fig.3 shows the main systems of effective dislocations in this case. It is shown that the effective dislocations create in the central area of the specimen the three-axial tensile stress. Microcrack nucleation begins as the resulting tensile stresses reach the material strength limit, even though the external compressive stresses are present. The main crack growth occurs in accordance with the plastic mechanism suggested above: the interrelation between the dislocations of various signs (See Table 1) squeezing into the pore leads to the pore steady growth both along the radius and along the axis within the whole of the area of the tensile stresses. So, even the materials of high plasticity whose uniaxial

elongation can exceed 50 per cent display fracture behaviour in cross-rolling at ϵ as small as 5 per cent. The study of the role of the different process parameters in the effective dislocation structure formation enables us to work out the optimum regimes of the material treatment.

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Table 1. Types of Crack Propagation Mechanisms

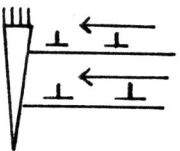
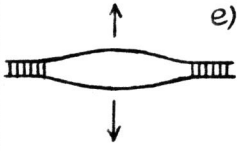
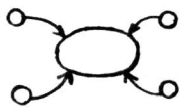

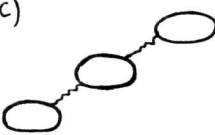
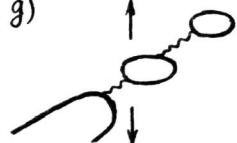
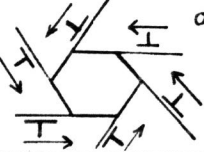
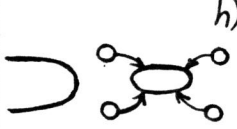
1.Type 2.Energy influx in the form	Scheme	1.Type 2.Energy losses in the form	Scheme
1 Brittle ^{2/} 2 Elastic energy of dislocations		1 Brittle 2 No energy losses	
1 Vacancy 2 Energy of vacancy formation		1 Quasi-brittle ^{3,4/} 2 Heat	
1 Statistic 2 Plastic deformation work energy		1 Ductile ^{3,4/} 2 Plastic deformation work and heat	
1 Plastic 2 Dislocation core energy		1 Ductile ^{5,6/} 2 Heat	



Fig1. Scheme of generation of effective dislocations near the surface



Fig3. Generation of effective dislocations during cross-rolling in cross (a) and not axial (b) sections



Fig2. Division of dislocations into dipole and effective ones